Characterization on Microstructure and Carbides in an Austenitic Hot-work Die Steel during ESR Solidification Process

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1. Introduction

The current work was undertaken to systematically examine the as-cast microstructure and carbides in a developed austenitic hot work die steel produced by conventional electroslag remelting (ESR) and continuous directional solidification of electroslag remelting (ESR-CDS). In addition, the growth pattern of carbides was also discussed. A combination of optical microscopy (OM) and scanning electron microscopy (SEM) were used to characterize microstructure and carbides. The segregation was analyzed using an original position analyzer (OPA) and electron probe microanalysis (EPMA). The electrolytically extracted carbides were analyzed by SEM and x-ray diffraction (XRD) to identify their three-dimensional microstructure and compositions. The microstructure of electroslag remelted austenitic die steel was composed of austenite matrix and primary carbides V₆C₇-type and Mo₂C-type. Compared with conventional ESR, ESR-CDS contributed to a finer as-cast microstructure, a smaller amount and smaller size of carbides in remelted steel. Meanwhile, the alloying elements segregation was reduced through ESR-CDS. The enrichment of carbide-forming elements was reduced through directional solidification of ESR, resulting in the change in the morphology of V-rich carbides from rod-like to lamellar-shaped. The hardness and V-notched impact energies of remelted ingot (produced by ESR-CDS) after heat treatment (solution temperature 1180°C for 2 hours, aging temperature 720°C for 2 hours) was increased by 3 to 5HRC and 4 to 6 J/cm² respectively, in comparison with that produced by conventional ESR.

KEY WORDS: electroslag remelting; microstructure; carbides; directional solidification; segregation; die steel.

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to obtain the better strength and toughness.\(^\text{1}\) The defects caused by large primary carbides are difficult to eliminate during forging and heat treatment stages. A well-known way for controlling primary carbides is to reduce element segregation during solidification. Several methods have been explored, including adding nucleating agent or alloying elements;\(^\text{5-6}\) mechanical or electromagnetic stirring;\(^\text{7-9}\) etc. However, the electroslag remelting (ESR) process is considered as a widely used secondary refining technology for producing high-quality special steel ingots. The ESR method can not only refine microstructure and homogenize component, but also greatly remove the non-metallic inclusions of electrode.\(^\text{10}\)

The solidification quality of ESR ingots is closely related to the depth of the metal pool. Some research\(^\text{11}\) showed that shallow and flat molten pool can shorten the local solidification time and restrict the carbide segregation. In respect to the traditional fixed-mould ESR process, the depth of metal pool was generally modified by these measures as follows: decreasing the remelting speed of electrodes;\(^\text{12-14}\) increasing the fill ratio of the electrode cross-sectional area to the mold cross-sectional area;\(^\text{15}\) increasing the appropriate weight of slag.\(^\text{16,17}\) However, the shape of the molten-metal pool changes gradually in the traditional fixed-mould ESR process, so it is difficult to achieve the critical balance between surface quality and internal quality. Moreover, remelting rate is not a linear relationship with the local solidification time. Therefore, the traditional fixed-mould ESR limits the ability to control carbide segregation in ingot. Shi et al.\(^\text{18}\) and Chumanov et al.\(^\text{9}\) reported that the rotation of a mold and consumable electrode respectively in ESR process can not only alleviate the segregation of carbides and reduce the size of carbides in the high speed steel, but also reduce inclusions and improve the surface quality of ingots. However, these measures above are feasible to achieve the shallow and flat molten pool of ESR in small section ingots, but they exhibit some limitations in large cross section ingots.

Base on cooling excitation theory, continuous directional solidification of electroslag remelting (ESR-CDS) was introduced in this investigation by applying secondary aerosol cooling water at 160 mm below mold when ingot drawing out of furnace. The liquid steel was cast into a rod, and then forged into two rods of 120 mm in diameter. Then the electrodes were remelted using conventional ESR and ESR-CDS technology for comparison, and the correspondingly produced as-cast ingots of 160 mm in diameter were sampled as E-1 and E-2, respectively. The remelting process was conducted in the argon gas atmosphere. The chemical compositions of remelted ingots E-1 and E-2 were determined by inductively coupled plasma optical emission spectrometer, and the results are shown in Table 1.

The schematic diagrams of the traditional fixed-mould ESR and ESR-CDS apparatus were illustrated in Figs. 1(a) and 1(b), respectively. In comparison with traditional ESR process, the improved measures in ESR-CDS process include these methods as follows: applying electric conductively water-copper mold with water-cooled baseplate continuously downward; spraying secondary aerosol cooling water at 160 mm below mold when ingot drawing out of the hot zone. The similar remelting speed of the two ingots were taken by adjusting the input voltage, current and other process parameters, and the process parameters were listed in Table 2.

The as-cast ingots E-1 and E-2 experience the same deformation condition through hot forging, then undergo the same heat treatment shown in Fig. 2. The heat treatment process mainly included solid solution strengthening treatment and aging precipitation strengthening treatment. The hardness after aging treatment was measured with a Rockwell hardness tester (THR-150D, HRC) and the value was taken the average of five points hardness with every sample. Meanwhile, impact test on Charpy V-notched type samples was performed for determining impact toughness with a JB30B impact test machine at room temperature.

### 2. Experimental

#### 2.1. Experimental Materials

The austenitic hot-work die steel was obtained by melting pure alloy ingredients in a 200 kg vacuum induction furnace. The liquid steel was cast into a rod, and then forged into two rods of 120 mm in diameter. Then the electrodes were remelted using conventional ESR and ESR-CDS technology for comparison, and the correspondingly produced as-cast ingots of 160 mm in diameter were sampled as E-1 and E-2, respectively. The remelting process was conducted in the argon gas atmosphere. The chemical compositions of remelted ingots E-1 and E-2 were determined by inductively coupled plasma optical emission spectrometer, and the results are shown in Table 1.

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#### 2.2. Microscopic Observation

The samples with the dimension of 15 mm×15 mm×12

| Table 1. Chemical composition of steel (wt%). |
|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|-----------------|
|                | C   | Si  | Mn  | Cr  | Mo  | V   | P   | S   | Fe        |
| E-1            | 0.70| 0.55| 14.95| 3.45| 1.57| 1.72| 0.0085| 0.0023| Bal.      |
| E-2            | 0.70| 0.54| 14.90| 3.53| 1.55| 1.73| 0.0088| 0.0021| Bal.      |

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mm were taken from the same position of as-cast ingots E-1 and E-2, respectively, at the top position of 1/4 of longitudinal ingots. The metallographic samples were analyzed by OM (LEICA DM2500M) and SEM (FEI MLA250, FEI, Hillsboro, OR, USA) equipped with EDS (XFlash 5030; Bruker, Germany), after grinding, polishing and etching with 6% nitric acid alcohol.

Carbon segregation with the size of 80 mm×80 mm in the transverse section of both ingots E-1 and E-2 from center to boundary was analyzed using an original position analyzer (OPA 200, NCS, China). OPA is a technology based on continuous excitation spark spectroscopy with two-dimensional scanning of specimens, high-speed signal acquisition of single spark discharge, and data analysis. The OPA instrument can scan large-sized specimen and generate signals reflecting both the concentrations and locations of the elements in the scanned areas.

2.3. Thermodynamical Calculate of Carbide Precipitation

The Thermo-Calc software with the TCFE7 database was employed to calculate solidification and carbides precipitated temperature sequence of austenitic hot-work die steel. The chemical composition input in this program is listed in Table 1. The simulation was performed under both the equilibrium condition and the non-equilibrium condition taking the Scheil model into account.

2.4. Carbides Observation in Three Dimensions

To observe the morphology of carbides with an original position analysis method in as-cast ingot, the metallographic samples were deeply etched for 1 min with the etchant (5 g FeCl₃+10 mL HNO₃+3 mL HCl+87 mL alcohol). To investigate the three-dimensional features of the carbides, the samples taken from E-1 and E-2 ingots, were machined into a rod of Ø15 mm×90 mm. Carbides were extracted from steel matrix in organic solution (methanol, tetramethylammonium chloride, glycerin, diethanol amine) by electrolysis. Some of the carbides were analyzed by XRD (Rigaku D_max-RB, Rigaku, Tokyo, Japan) to confirm the types, and some carbides were observed by SEM for the three-dimensional morphology. The composition of carbides with different morphologies were measured by EDS.

3. Results and Discussion

3.1. Macrosegregation Analysis

The 2D contour maps of carbon distribution scanned by OPA in the same transverse section of ingots E-1 and E-2 were shown in Fig. 3. The degree of the carbon contents in ingots was represented by different colors, which increases with the transition of colors change from blue to red. It can be seen by comparing Figs. 3(a) and 3(b) that more uniform distribution of carbon and less segregation were achieved in the case of the ingot E-2 subjected to the ESR-CDS process, and which had been described in detail in previous work.25)

Fig. 1. Schematic diagrams of two types of electroslag remelting: (a) ESR; (b) ESR-CDS.

<table>
<thead>
<tr>
<th>Process parameters</th>
<th>ESR</th>
<th>ESR-CDS</th>
</tr>
</thead>
<tbody>
<tr>
<td>Diameter of mold, mm</td>
<td>160</td>
<td>160</td>
</tr>
<tr>
<td>Diameter of electrode, mm</td>
<td>120</td>
<td>120</td>
</tr>
<tr>
<td>Voltage, V</td>
<td>38</td>
<td>38</td>
</tr>
<tr>
<td>Electricity, A</td>
<td>2 400</td>
<td>2 100</td>
</tr>
<tr>
<td>Weight of slag, kg</td>
<td>6</td>
<td>5</td>
</tr>
<tr>
<td>Melting rate, kg/min</td>
<td>1.65</td>
<td>1.6</td>
</tr>
</tbody>
</table>

Fig. 2. Schematic diagram of the heat treatment process in the experiment.

It is well known that the elements Cr, Mo and V are strong former of carbides, and the segregation of these elements are beneficial to the formation of carbides during solidification. Thus, to control the characteristics (i.e. amount, size, morphology and distribution) of carbides precipitated from residual liquid during solidification, the segregation of carbon and carbides formed elements should be alleviated. Therefore, the segregation of elements in residual liquid during solidification in ESR and ESR-CDS...
process were detected through EPMA, and the results were shown in Fig. 4.

It can be seen from Figs. 4(c) and 4(d) that fluctuation degree of scanning line of elements in ingot E-1 is much more intense than that in ingot E-2, especially for elements V and Mn. Fluctuation degree of scanning line means the difference of the elements mass percent distributed between dendrite arm and interdendritic area. It can be concluded that the dendritic segregation of elements in ingot E-2 was alleviated through ESR-CDS process compared to that in ingot E-1 produced through ESR process. Meanwhile, the distance of peak of scanning line represents the dendritic arm spacing. Combination of SEM images in Figs. 4(a), 4(b) and scanning lines in Figs. 4(c), 4(d), the dendritic arm spacing in ingot E-2 is much smaller than that in ingot E-1. It can be concluded that dendritic microstructure could be refined through ESR-CDS process in comparison with ESR process.

The study related to the dendritic arm spacing calculation was presented in subsequent section. In addition, the precipitation behavior of carbides during solidification is related to the enrichment degree of solute atoms. The concentration degree of solute atoms in residual liquid located interdendritic region in ingot E-2 is lower than that in ingot E-1 (seen in Fig. 4). Thus, the correspondingly supersaturation of carbon in the remaining liquid at the last stage solidification is lower in ingot E-1, which also leads to the relatively lower growth rate of the carbides. And the thermodynamic condition of precipitated carbides during solidification in ESR-CDS process should be inhibited due to uniform distribution of solute atoms. Lastly, the solidification rate in ESR-CDS process is greatly larger than that in ESR process, which results to insufficient time to the diffusion of solute atoms to satisfy the further growth of carbides. Therefore, the size of carbides could be smaller through ESR-CDS. This suppose is obviously presented in Figs. 4(a) and 4(b). The amount of precipitated carbides in ingot E-1 is much more than that in ingot E-2, as well as the size of precipitated carbides in ingot E-1 is much larger than that in ingot E-2. The detail characteristics of precipitated carbides were described in the subsequent study presented in this paper.

3.2. Microstructure in As-cast Ingot

Figure 5 shows the solidification structures with longitudinal section at different transverse section positions of the ingots from edge to center produced through ESR and ESR-CDS process. The secondary dendrite arm spacings were measured in a statistical averaging way to prevent randomness and the correspondingly results were shown in Fig. 6. The results indicated that the secondary arm spacing in ingot E-2 were much smaller than that in ingot E-1,
which may be caused by factor that the solidification rate in ESR-CDS was much larger than that in ESR.

It has been reported that the decrease in the dendrite arm spacing with increase in the solidification rate can be explained by the fact that there is insufficient time available for the diffusion of solute. The local solidification rate could be calculated by Eqs. (1)–(3) as follows:

\[
\lambda_{SDAS} = 143.9 \cdot C_R^{0.3616} \cdot [\%C]^{(0.5501-1.996[\%C])} \quad (1)
\]

\[
t_f = \frac{T_L - T_S}{C_R} \quad (2)
\]

\[
[\%C] > 0.15 \quad (3)
\]

Where, [\%C] is the carbon content in steel (mass pct), \(\lambda_{SDAS}\) is the secondary arm spacing (\(\mu m\)), \(t_f\) is the local solidification time (s), \(T_L\) and \(T_S\) are liquidus and solidus temperatures (K), \(C_R\) is the solidification rate (K·s\(^{-1}\)).

The correspondingly liquidus and solidus temperature calculated by using the Thermo-Calc thermodynamic software are 1653 K and 1493 K, respectively. The microstructure observation in Fig. 5 and the statistic value of secondary dendrite arm spacing in Fig. 6, the value of secondary dendrite arm spacing at position of 1/2 radius was used to calculate the solidification rate during solidification in ESR and ESR-CDS process. For as-cast ingots E-1 and E-2, the average secondary dendrite arm spacing for representation are 67 and 36 \(\mu m\); thus, the estimated average local solidification rates are 5.77 and 31.14 K·s\(^{-1}\), respectively. In conclusion, the local solidification rate in ESR-CDS process is greatly larger than that in ESR process, which attributes to the reason of the decreasing of dendritic arm spacing in ingot E-2 produced through ESR-CDS.

The solidification rate is not only one critical factor to control the morphology and growth orientation of dendrites, but also an essential factor to reform the features (i.e., amount, size, distribution and morphology) of precipitated carbides during solidification. The precipitated carbides in ingots E-1 and E-2 were characterized by image analysis software and correspondingly statistic results were shown in Fig. 7. The characteristic parameters included in Fig. 7 are as follows: the number of carbides per unit area (number density/mm\(^2\)); the total areas of carbides per unit area (%); the average size of carbides (\(\mu m\)); the area of carbides; volume fraction (%). Comparison in characterization of carbides between ingot E-1 and E-2 shown in Fig. 7, the number density decreases slightly, but the other features (i.e., area ratio, average size and volume fraction) of carbides decrease sharply in ingot E-2 produced through ESR-CDS process. This decreasing tendency of characterizations of carbides indicated that the microstructure with small size and dispersed distribution of
carbides can be obtained by ESR-CDS process.

3.3. Morphologies of Precipitated Carbides

To understand the effect of directional solidification in electroslag remelting process on the formation and growth behavior of precipitated carbides, the two-dimensional and three-dimensional structure of carbides were extracted to detect the feature in morphology of carbides. The morphologies of precipitated carbides were shown in Figs. 8, 9 and 10, and the composition detected by EDS was listed in Table 3. Combination of morphologies of carbides (Figs. 8 and 9) with composition of carbides (Table 3), it can be known that carbides extracted from ingot E-1 and E-2 could be divided into two types. One is gray MC-type carbides which were vanadium-rich containing a certain amount of Cr and Mo elements (Fig. 8); the other is white fibrous-like M₂C-type carbides which were molybdenum-rich containing a certain amount of Mn, Cr and V elements, and developed sparsely (Fig. 9). To confirm the types of precipitated carbides in ingot E-1 and E-2 during solidification, the extracted carbides powder was analyzed through XRD. The XRD results shown in Fig. 11 indicated that the MC-type and M₂C-type carbides were V₅C₇ and Mo₂C, respectively.

The three-dimensional morphology of MC-type carbides shown in Fig. 8(c) was short rod-shaped, while that in Fig. 8(d) was disc-shaped with multi-angles.¹¹) Meanwhile, from the in-situ observation of carbides morphology in deep-etched samples shown in Figs. 8(a) and 8(b), it is obviously known that the MC-type carbides precipitated along preexisted austenitic grain boundaries. The morphology of MC-type carbides was influenced by the solidification rate and

![Fig. 8](image). Morphology of MC-type carbides: (a), (c) ingot E-1; (b), (d) ingot E-2. (Online version in color.)

![Fig. 9](image). Morphology of M₂C-type carbides: (a), (c) ingot E-1; (b), (d) ingot E-2. (Online version in color.)

![Fig. 10](image). Mix morphology of MC and M₂C-type carbides: (a), (c) ingot E-1; (b), (d) ingot E-2. (Online version in color.)

<table>
<thead>
<tr>
<th>Ingot</th>
<th>Precipitated Carbides</th>
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<tr>
<td></td>
<td>C</td>
<td>Fe</td>
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<tr>
<td>E-1</td>
<td>MC-type</td>
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</tr>
<tr>
<td></td>
<td>M₂C-type</td>
<td>13.68</td>
</tr>
<tr>
<td>E-2</td>
<td>MC-type</td>
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</tr>
<tr>
<td></td>
<td>M₂C-type</td>
<td>14.82</td>
</tr>
</tbody>
</table>

![Fig. 11](image). XRD pattern of the carbides powder.
parallel. It can be known that the M2C-type carbides exhibits irregular to each other, they were both fibrous-shapes. While from ingot E-1 and E-2 shown in Figs. 9(a) and 9(b) were similar in morphology of MC-type carbides produced in ESR and ESR-CDS process. In addition, the solidification rate in ESR-CDS process is greatly larger than that in ESR process may be another factor to control the morphology of MC-type carbides.

The two-dimensional morphology of M2C-type carbides in ingot E-1 and E-2 shown in Figs. 9(a) and 9(b) were similar to each other, they were both fibrous-shapes. While from the three-dimensional morphologies shown in Figs. 9(c) and 9(d), they were lamellar and long strip shape arranged in parallel. It can be known that the MC-type carbides exhibits the dendritic structure. This dendritic morphology of M2C-type carbides may be attributed to the temperature difference between the surface and center of the ingot when M2C-type carbides began to precipitate during solidification. Figure 9 shows that a midplane of M2C-type carbides forms firstly along the direction of the heat flow, then a series of branches grow from it all around. When the branches of MC-type carbides meet each other or come across to austenite matrix and the branches of MC-type carbides (shown in Fig. 10), M2C-type carbides stops growing.

The mix morphology of MC and M2C-type carbides were shown in Fig. 10, this indicated that the precipitation of MC-type carbides occurred accompanying with the formation of M2C-type carbides. The reason for the mix morphology of MC and M2C-type carbides could be explained as follows. During the solidification in electroslag remelting process, the solute atoms from each phase in the eutectic mixture were exhausted to the liquid and depart from the front of solid-liquid interface by diffusion, then deposit on the other phase finally. MC-type carbides need plenty of alloy element V atoms to ensure its growing, and alloy element V atoms in sounding region spreads to where MC-type carbides exist and the alloy elements Mo and Cr atoms are left around MC-type carbides. Meanwhile, the eutectic M2C-type carbides needs plenty of alloy elements Mo and Cr during its growth and alloy element V is left around M2C-type carbides. The alloy elements Mo and Cr atoms diffused from the region of MC-type carbides are employed for the growth of M2C-type carbides, and the alloy element V transfers to the place where MC-type carbides grow up. Therefore, the different eutectic carbides mix together and grow corporately can be observed in Fig. 10.

3.4. Growth Pattern of Precipitated Carbides

To illustrate the precipitation behavior of eutectic carbides during solidification, the equilibrium and non-equilibrium phase diagram of austenitic hot-work die steel were calculated by Thermo-Calc software. The corresponding results were shown in Fig. 12.

It can be seen from Fig. 12(a) that the temperature of precipitated carbides (i.e., MC, M2C, M7C3 and M23C6) in sequence during equilibrium solidification condition were 1 380, 1 210, 920, 835 and 660°C, respectively. However, during actual solidification process, the segregation of elements and super cooling phenomenon are inevitable to occur. Thus, the temperatures of precipitated carbides have a certain distinction. Figure 12(b) shows that MC and M2C carbides precipitated from the residual liquid with mass fraction of solid from 0 to 1. The correspondingly eutectic reactions as follows: L→γ+MC occurs at 1 210°C (fL = 0.86) and L→γ+MC+M2C occurs at 1 160°C (fL = 0.95). However, M3C3 and M23C6-type carbides cannot be found in microstructure, which may be attributed to rapid solidification resulting to precipitation of carbides from austenite phase scarcely. Thus, the microstructure consists of austenite, eutectic MC and M2C-type carbides.

The precipitation temperatures of MC and M2C-type carbides indirectly verify the explanation about the morphology of carbides affected through ESR-CDS process. The precipitation temperature of MC-type carbides is higher than that of M2C-type carbides, which results to the influence by local concentration of solute atoms and local solidification rate on morphology of MC-type carbides is greatly larger than that on morphology of M2C-type carbides. Therefore, under the influence of directional solidification in ESR-CDS process.
the morphology of MC-type carbides was changed, while that of M2C-type carbides was similar in comparison with ESR process. In addition, the eutectic reactions of precipitated carbides were in agreement with the illustration about formation behavior of carbides in previous section.

The morphology of eutectic MC-type carbides has been generally investigated by previous research, while the growth pattern of carbides was rarely reported. It was known that MC-type carbides take an FCC structure with the most stable plane of \{111\}. The octahedron consisting of \{111\} faces is therefore the equilibrium shape. The growth patterns of MC-type carbides present dendritic morphology, the correspondingly morphology were shown in Fig. 13. The difference in growth patterns of MC-type carbides between ESR and ESR-CDS process is dendritic morphology with rod shaped (seen Figs. 13(a), 13(b), 13(c)) and dendritic morphology with lamellar shaped (see Figs. 13(d), 13(e), 13(f)). The reason for dendritic carbides is the carbides grow along the <001> directions more quickly than other directions for the least atoms in the \{001\} faces of the FCC crystal. The secondary and tertiary dendritic arms were observed in Fig. 13. The schematic diagram of growth patterns of MC-type carbides was shown in Fig. 14. The growth steps were illustrated at the tips of the carbides, the development and connection of these secondary and tertiary dendritic arms form the flake-shaped MC-type carbides in the end.

![Fig. 13. The morphologies of MC-type carbides in ingots E-1 and E-2: (a), (b), (c) ingot E-1; (d), (e), (f) ingot E-2.](image1)

![Fig. 14. The schematic diagram of growth patterns of MC-type carbides.](image2)

3.5. Effect of Heat Treating on Microstructure and Mechanical Properties

The SEM microstructure photographs of ingots E-1 and E-2 under aging treatment state were shown in Fig. 15. And the corresponding hardness is 43 HRC and 48 HRC, respectively. It is clearly that the hardness at aging heat treatment state is greatly affected by the extent of solid solution hardening by carbon and alloying elements, as well as the precipitation of finer secondary carbides from austenite matrix. It can be seen from Fig. 15 that the primary coarsening MC-type carbides in Fig. 15(b) were completely dissolved through solid solution heat treatment, while that in Fig. 15(a) were incompletely dissolved. Moreover, microstructure with the precipitation of secondary carbides in Fig. 15(b) obviously distributed more uniform and presented finer size in comparison with that in Fig. 15(a). In addition, the M2C-type carbides in both ingot E-1 and E-2 were dissolved through heat treatment. The more elements dissolved from carbides, the more elements infused into austenite matrix. Therefore, the higher carbon and alloying elements dissolved into austenite matrix. Meanwhile, the finer size and disperse distribution of secondary carbides were another factor to improve the hardness in Fig. 15(b).

![Fig. 15. SEM microstructure photographs of ingots E-1 and E-2 under aging treatment state: (a) ingot E-1; (b) ingot E-2. (Online version in color.)](image3)

The SEM fracture graphics of Charpy impact test performed at room temperature to evaluate toughness of the two ingots E-1 and E-2 under aging treatment state were presented in Fig. 16. The values of impact energy were 9.5 J cm\(^{-2}\) and 13.6 J cm\(^{-2}\) for ingots E-1 and E-2 respectively.
under aging treatment state. The SEM analysis on fracture surface indicates that, the fracture characteristic of ingot E-1 is quasi-cleavage in Fig. 16(a) after aging heat treatment, while for ingot E-2, it is quasi-cleavage fracture with mass small dimples in Fig. 16(b). The basic characteristics of quasi-cleavage fracture are small cracked grain and tear ridges. The cracks generally derive from area of stress concentration such as inclusions and carbides. The fracture surface presented much more amounts of carbides and larger size of carbides in Fig. 16(a) than that in Fig. 16(b). Thus, it is suggested that superior toughness is associated with minimization of large primary carbides. The undissolved primary carbides have a greater effect on toughness than that of carbides precipitated upon aging heat treatment. Combination microstructure photographs in Fig. 15 with fracture photographs in Fig. 16, it can be concluded that the undissolved large primary MC-type carbides in ingot E-1 mainly reduce the impact toughness at aging heat treatment state in comparison with ingot E-2.

The reasons for the values of hardness and toughness in ingot E-2 are higher than that in ingot E-1 at aging heat treatment state can be explained as follows. Firstly, the microstructure of dendrites was refined and composition of alloying elements was homogenized in ingot E-2 produced through ESR-CDS. Moreover, the refinement of carbides with small size, less quantity and disperse distribution in ingot E-2 could be obtained through ESR-CDS. The reasons for the values of hardness and toughness in ingot E-2 are higher than that in ingot E-1 at aging heat treatment state can be explained as follows. Firstly, the microstructure of dendrites was refined and composition of alloying elements was homogenized in ingot E-2 produced through ESR-CDS. Moreover, the refinement of carbides with small size, less quantity and disperse distribution in ingot E-2 could be obtained through ESR-CDS.

(3) The morphologies of MC-type carbides were changed from rod-shaped to lamellar-shaped under directional solidification in ESR-CDS process. However, the morphologies of M2C-type carbides in ingots E-1 and E-2 have similar microstructure. The growth pattern of MC-type carbides presents dendritic morphology. The difference in growth patterns of MC-type carbides between ESR and ESR-CDS process is dendritic morphology with rod shaped in ESR process and dendritic morphology with lamellar shaped in ESR-CDS process.

(4) The hardness and toughness of austenitic hot-work die steel after aging heat treatment were all remarkable improved produced through ESR-CDS process. The hardness of ingot E-2 was 48 HRC, while that of ingot E-1 was 43 HRC after aging heat treatment, which has been improved 5 HRC. Meanwhile, the toughness of ingot E-2 was 13.6 J cm\(^{-1}\), while that of ingot E-1 was 9.5 J cm\(^{-1}\), which has been improved 4.1 J cm\(^{-1}\).

Acknowledgments
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